1. Introduction

In the past, hydrogen induced cold cracking (HICC) was observed most commonly in the heat-affected zones (HAZ) of high strength steels. Along with the improved resistance to HAZ cracking achieved in modern high strength steels, however, weld metal becomes more prone to suffer from cold cracking than the HAZ of base metal. In particular, weld metal cracking has been reported to become a major limiting factor in the preheat-free application of HSLA steels, which had been developed for substituting HY steels and to eliminate the preheating requirement. Accordingly the phenomenon of HICC in high strength weld metal has received some attention recently but the factors controlling cold cracks in weld metal are not well understood.

Cold cracks formed in weld metal have been reported by Flanigan et al. for the first time. They observed these cracks in the bead-on-plate E6010 weld beads subjected to quenching immediately after welding. They were found in the form of micro-cracks lying in planes perpendicular to the axis of the weld and were noticed to be concentrated in the interior of the weld metal, i.e. not open to the bead surface. As these cracks were not developed in the low-hydrogen type E6016 welds they could conclude that hydrogen was a factor of prime importance for the formation of micro-cracks in weld metal. Afterwards, much work has been done using single pass weld techniques like the constant load rupture test, the Y-groove Tekken test, the tensile restraint cracking test and the gapped bead-on-plate test. The principal variables which have been found to be important were strength of weld metal, microstructure, hydrogen level, restraint, and the weld cooling rate. As would be expected from consideration of HAZ cracking, the risk of cracking has been found to increase universally with the increase of hydrogen, restraint and the cooling rate, and generally with the increase of weld metal strength. In case of strength, however, Hart reported that, at a low hydrogen content and in the weld of fine microstructure and/or Ni addition, the risk of cracking decreased even with the increase of weld metal strength.

Those studies performed with single pass weld have provided much valuable information, but they were of little assistance in the provision of procedural guidance for multipass welds in which not only the hydrogen content but the residual stress accumulate with building up the weld passes. Besides, the weld microstructure formed in multipass welds, as being affected by subsequent thermal cycles,

© 2003 ISIJ 706

Effect of Microstructural Variation on Weld Metal Cold Cracking of HSLA-100 Steel

H. J. KIM and B. Y. KANG

Advanced Joining & Welding Research Team, KITECH 35-3 HongChonRi IbjangMyun, ChonAnSi, Korea.
E-mail: kimhj@kitech.re.kr, kanbo@kitech.re.kr

(Received on October 10, 2002; accepted in final form on November 21, 2002)
is not relevant to that of single pass weld. Accordingly, the condition of three major factors, i.e., susceptible microstructure, diffusible hydrogen and residual stress, contributing to cold cracks in a multi-pass weldment is far different from those in single pass one. Several studies, mostly in Japan, have been performed on the risk of weld metal cracking in the multipass weld.\textsuperscript{13-15} In Refs. (13) and (14), investigators have deposited weld beads into a highly restraint V-groove joint with varying the weld metal strength, the hydrogen content and the height of weld metal. The preheat temperature necessary to prevent weld metal cracking was determined as a function of those variables and then proposed the following empirical equation, Eq. (1);

$$T(\text{°C}) = A R_m + B \log[H] + C h_w + D$$

where $R_m$ is the weld metal tensile strength, [H] is the weld metal diffusible hydrogen content, $h_w$ is the weld metal height, and $A$, $B$, $C$ and $D$ are constants. Unlike the various equations currently available for predicting the preheating temperature necessary to prevent HAZ cracking,\textsuperscript{16,17} above equation does not contain a term like carbon equivalent but is showing a strong linear relationship with tensile strength of weld metal ($R_m$). If this is the case the multipass weld might have different resistance to HICC from location to location as it has a large variation in microhardness throughout the weldment, especially in high strength weld metals. Previously Moon et al.\textsuperscript{18} measured the microhardness for multipass ultra-low carbon weld metal deposited in HSLA-100 steel and reported a variation of microhardness in the range between 260–320 Hz. The lowest hardness was recorded in the localized band that located in the outer portions of HAZ of weld bead. As this band was revealed in bright contrast in their optical micrographs they called it as ‘white band’. The maximum hardness, on the other hand, was recorded in the midway between the fusion boundaries and white band and thus the distance between the maximum and minimum hardness was less than 1 mm. It means that the most susceptible microstructure is located just adjacent to the least susceptible one. Under such a situation, it would be very interesting to figure out the effect of local variation in microhardness on crack susceptibility and its propagation. This is the first purpose of this investigation.

Traditionally, as described above, the microhardness has been regarded as a quantitative measure to evaluate the microstructural susceptibility to cold cracking not only in HAZ but also in the weld metal.\textsuperscript{15} In high strength steels, however, it has been well documented that hydrogen embrittlement is strongly affected by microstructure.\textsuperscript{19} For weld metal, it was also reported that higher hardness was not always corresponding to higher susceptibility and showed a combination of high strength and less susceptibility could be achieved through microstructural modification.\textsuperscript{20,21} In other words, the susceptibility could be different depending on the weld microstructure even at the same hardness or strength level. Accordingly, the second purpose of this paper is to figure out the microstructural constituent that would be most susceptible to HICC in high strength, multipass weld metal.

2. Experimental Procedure

After machining to have V-groove with 60-deg included angle, HSLA-100 steel plates of 25 mm thickness were welded in a highly restraint condition. As shown in Fig. 1, fillet welds were made to restrain the plates and to prevent the distortion. The experimental variable was the preheating temperature, which was set at 75, 100, and 125°C. The temperature of the test assembly was maintained within a range of ±5°C from the designated temperature during welding. The first two passes were made with AWS ER120S by gas–metal arc welding (GMAW) process, and the last of weld passes were made with submerged arc welding (SAW) process using a commercial combination of wire and flux. The chemical compositions of base metal, SA welding wire and deposited weld metal are provided in Table 1. Prior to test welding, the diffusible hydrogen content of SA weld was measured to be 7.1 ml/100 g with gas chromatographic method in accordance with WES 1003.

Welding parameters are given in Table 2. All welding was performed with AC power supply. Transverse cross section taken from the completed weldment is shown in Fig. 2(a). Groove was not filled up completely in order to prevent the possible release of residual stress in the portion of reinforcement. A total of 11 weld beads were made including first two GMA weld beads. In Fig. 2(a), it is readily apparent that there are two different sets of boundaries. One was the fusion boundary of each bead and it gave clear information on build-up sequence of weld beads. This sequence is illustrated in Fig. 2(b), where the fusion boundaries are presented with solid lines. The other boundaries were rather like bands surrounding each weld beads with some distance from fusion boundaries. As mentioned earlier, Moon et al.\textsuperscript{18} also noted this sort of bands which were

\begin{table}[ht]
\centering
\caption{Chemical compositions.}
\begin{tabular}{lcccccccc}
\hline
 & C & Si & Mn & P & S & Cu & Ni & Cr & Mo \\
\hline
 Base Metal  \\
 (HSLA-100) & 0.05 & 0.21 & 0.50 & 0.008 & 0.002 & 1.29 & 3.50 & 0.59 & 0.48 \\
 Welding Wire & 0.14 & 0.08 & 1.72 & 0.007 & 0.007 & 0.04 & 2.13 & 0.35 & 0.57 \\
 Weld Metal & 0.11 & 0.10 & 1.67 & 0.010 & 0.009 & 0.20 & 2.29 & 0.36 & 0.57 \\
\hline
\end{tabular}
\end{table}

\begin{table}[ht]
\centering
\caption{Welding conditions.}
\begin{tabular}{lcccc}
\hline
Wire Diameter & Welding Variables & Heat input & Preheating Temperature \\
 & Current & Voltage & Speed & \ & & \\
4.0mm & 420A & 28V & 400mm/min & 17.0kJ/cm & 75, 100, 125°C \\
\hline
\end{tabular}
\end{table}
developed with bright contrast in their optical micrographs, and called them as ‘white bands’. However, in the present study, it was found that these bands could be revealed either in bright or in dark contrast depending on the etching condition. Those bands shown in Fig. 2(a) are the ones revealed in dark contrast. As these bands are corresponding to the softest region, which has been reported by Moon et al.\textsuperscript{18)} and will be confirmed later in this study, they will be called as ‘tempered bands’ in the present paper. In Fig. 2(b), the profile of these bands is illustrated with dotted lines.

Microhardness maps across the base metal, heat affected zone and fusion zone were constructed using a diamond hardness tester with a 0.5-kg load. After polishing away the etched surface slightly, indentations were made with 1 or 0.5 mm spacing across the areas pre-determined. The hardness results were then interpolated using photoshop software on a computer and transformed into a map with black-and-white contrast. Detailed microstructural analysis was performed with optical microscopy. For getting a better image, several etchants were tried to delineate the transformation products and a solution of 1\% nital gave a best result.

After the groove joints were filled, the welded plates were left for more than 10 d in the open air and then sectioned in two different directions as shown in Fig. 2(b). Before sectioning, the welded joint of 20 mm in length was inspected with scanning acoustic microscope (SAM) to identify the exact location of hidden cracks and to characterize their two-dimensional features with respect to the variations in microstructure and microhardness. Based on the SAM information, the cutting line for longitudinal sectioning was determined crossing the crack image. ‘L’ surface of Fig. 2(b) was the one made for such a purpose. In addition, the top surface of welded plate was successively ground to examine the weld metal cracks from the top surface, which is marked as ‘T’ in Fig. 2(b).

3. Results and Discussion

3.1. Weld Microstructure

From the macrograph shown in Fig. 2(a), it can be realized that the present weld has a significant variation in microstructure. In order to characterize the local variation in microstructure, the optical specimen was taken from the final pass (11th bead) region and examined in detail. Figure 3 shows the optical micrographs taken from a specimen made with a preheating temperature of 125°C. As shown in Fig. 3(a), HAZ of HSLA-100 base metal (HAZ-BM) has a fully martensitic structure consisting of well-developed prior austenite grains, each of which is divided with several packets. In contrast, the weld metal has a structure of elon-
gated columnar grains containing a mixture of martensite and lath ferrite as shown in Fig. 3(b). When the preheating temperature was decreased, the volume fraction of martensite increased at the expense of lath ferrite, while grain boundary ferrite phase was barely formed. The distinctive features of lath ferrite have been reported by Fonda et al. for the ultra-low carbon weld metal in detail.

In the HAZ of weld metal (HAZ-WM), i.e. the part of the region located between the fusion boundary and tempered band, the elongated as-deposited grains were recrystallized by the subsequent weld thermal cycle(s) above the austenite transformation temperature, resulting in the equiaxed grains containing a mixture of martensite and lath ferrite, as shown in Fig. 3(c). Therefore these regions were called as ‘recrystallized HAZ-WM’ in this study and are illustrated with light gray contrast in Fig. 2(b). The regions of recrystallized HAZ-WM are surrounded with tempered bands. Crossing over the tempered band, no recrystallization took place and thus the columnar grain structure was preserved. Such regions having a structure of columnar grains are presented with dark gray color in Fig. 2(b). A part of these regions was reasonably expected to be tempered to a decreasing extent as getting away from the tempered band. However it was fail to find any indication of tempering within the limited resolution of optical microscopy. Further study in transmission electron microscopy (TEM) will be required.

3.2. Microhardness

Figure 4 shows the macrograph taken from the transverse cross-section of welded joint made with a preheating temperature of 100°C, the corresponding microhardness map constructed with a spacing of 1 mm and the overlapping image of those two figures. Figure 4(a) shows the similar patterns of fusion boundaries and black (tempered) bands as was shown in Fig. 2(a). The result of microhardness measurement performed on the region marked by a box is shown in Fig. 4(b). Macroscopically the upper part of weld metal is much higher in hardness than the lower region. This suggests a significant tempering took place in the lower beads with building up the weld passes. Figure 4(b) also shows that there is a local variation in microhardness within the weld beads deposited in the top layer. In order to directly correlate the microhardness variations to macrostructural features, including the as-deposited weld bead (11th bead), the recrystallized HAZ-WM, tempered band and less tempered HAZ-WM, the microhardness map was overlaid onto the macrograph to construct the overlapped image as shown in Fig. 4(c). Close examination of Fig. 4(c) were showed substantially low hardness values were recorded following the black bands that revealed clearly in the upper part of macrograph. In general, the hardness of last-pass weld metal is very high around 350 Hv extending across the recrystallized HAZ-WM, and then drops sharply to a minimum value near the black band. After passing the black band the hardness increases back to the as-deposited hardness. This result indicates that the black bands are the regions reheated and experienced a tempering cycle up to just below the austenitizing temperature. Based on this, these bands were called as ‘tempered bands’ in this paper. More details on hardness profile across the tempered bands will be described later.

3.3. Distribution of Cold Cracks

The prime interest of present investigation was to figure out the effect of local variation in microstructure and microhardness on the formation of cold cracks in multipass weld metal. As the weld under investigation has large variations both in microstructure (Fig. 3) and microhardness (Fig. 4(b)), it was expected to have a different degree of sus-
ceptibility to HICC from location to location. Accordingly it was aimed to find out the most preferred sites for cold cracks to be formed.

After the completion of welding and the elapsed time of 10 d, a visual inspection was performed on the as-welded plates and confirmed many surface cracks developed in transverse direction. All of these surface cracks were limited within the weld metal. In the longitudinal sections, these cracks were shown to propagate in the vertical direction up to the surface. It was also noted that there were numerous short cracks densely populated inside the weld metal, not open to the surface. To figure out the exact location of those hidden cracks present in the interior of weld metal, scanning acoustic microscope (SAM) was employed on the weld specimen which did not contain surface cracks. Figures 5(a) and 5(b) show the two-dimensional SAM image of hidden cracks and the corresponding macrograph overlapped with SAM image, respectively. This result demonstrates that cracks are absent in the last pass (11th bead) while most of cracks are populated in the adjacent beads (10th and 9th beads) and some of them are in the underneath beads (7th and 8th beads). It is also clear the cracks are in general limited in the areas of recrystallized HAZ-WM, which are surrounded by the tempered bands and the fusion boundaries. This direct correlation between the macrostructure and the crack locations was then used to determine the cutting lines for detailed analysis of cold cracks with respect to microstructural variation.

Based on the SAM result, the welded plates were cut in longitudinal direction crossing 10th and 7th beads (L-surface shown in Fig. 2(b)) in order to expose hidden cracks as many as possible. Typical examples of hidden cracks observed in the longitudinal sections are shown in Fig. 6. As demonstrated in these figures, there are a lot of hidden cracks exposed in these surfaces. One of the distinctive features is that they are all lined-up in two different levels, i.e. the cracks formed in the upper level follows the 10th bead and those in the lower level follows the 7th bead. The number of cracks is larger in the 10th bead that is located in the final layer than that in 7th bead. This result is not in agreement with that of Takahashi et al. They claimed that transverse cracks initiate in the weld beads just below the final layer and gradually propagate both to top and bottom surfaces. In this study, however, the weld beads in top layer turned out to be more preferential sites for the cold cracks to be developed. In extreme cases, the cracking reaches the surface of the weld metal resulting in long open cracks. Another interesting feature noticed in Fig. 6 was that the hidden cracks were appeared hardly to penetrate into the recrystallized region of weld metal (recrystallized HAZ-WM). This fact becomes quite clear in the high magnification optical micrograph shown in Fig. 7(a), which was taken at the tip of crack ‘A’ in Fig. 6(c). As shown in this figure, this crack stops its propagation right at the fusion boundary and thus limits only in the region of columnar grains. Further to this, as shown in Fig. 6(b), this crack tends to propagate along the columnar grain boundaries suggesting that this microstructure is more susceptible to cold cracking than the recrystallized HAZ-WM structure even though they are similar in hardness (Fig. 4).
3.4. Effect of Preheating Temperature

The apparent density of cracks in these sections was found to decrease with the increase of preheating temperature. **Figure 8** shows the quantitative evaluation of crack density expressed as a number of cracks per 50 mm in length. At preheating temperature of 75°C, there were 18 cracks along the 10th bead and 6 cracks along the 7th bead. Even with a preheating temperature of 125°C, there exists considerable number of cracks not only in the 10th bead but also in the 7th bead. In the actual fabrication, a post heating treatment at 200°C for 1 h was employed and was found to be inevitable for getting a crack-free weld. **Table 3** shows the mechanical properties obtained from the crack-free weld metal made with preheating temperature of over 125°C and the post heating treatment. The yield strength reached near 800 MPa and Charpy impact energy at −50°C was about 70 J.

3.5. Effect of Microhardness and Microstructure

The SAM image shown in Fig. 6(c) demonstrated that most of cracks were located in the two beads made just prior to the final pass and it appeared that these cracks hardly penetrate into the heat-affected region of weld metal (HAZ-WM). This fact became clear in the macrographs (Fig. 6) and micrographs (Fig. 7) taken in the longitudinal cross-sections and suggested the importance of microstructure in controlling HIC in weld metal. In order to confirm this feature more clearly and to clarify the effect of microhardness, the welded coupon that has been used for SAM examination was ground from the surface to reveal the T-surface, which is crossing the 11th, 10th and 9th beads as marked in Fig. 2(b). After etching this surface and the cross-section remained, the macrographs shown in **Fig. 9** were obtained. In the top surface (Fig. 9(a)), not only the fusion boundaries but also the tempered bands are revealed very clearly and are well matching with their locations represented in the cross-section (Fig. 9(b)). Near the center of the 10th and 9th beads, tempered bands are present which

![Fig. 7. Micrographs showing (a) crack termination at fusion boundary and (b) crack propagation along columnar grain boundary.](image)

![Fig. 8. Effect of preheating temperature on crack density.](image)

![Fig. 9. Typical cracks exposed on the ground top surface: (a) macrograph showing the preferential regions for crack development and (b) macrograph of cross-section matched with top surface.](image)

<table>
<thead>
<tr>
<th>Yield Strength</th>
<th>Tensile Strength</th>
<th>Elongation</th>
<th>R.A.</th>
<th>C_{at -50°C}</th>
</tr>
</thead>
<tbody>
<tr>
<td>797 MPa</td>
<td>890 MPa</td>
<td>15.8%</td>
<td>56.6%</td>
<td>69 J</td>
</tr>
</tbody>
</table>

![Table 3. Mechanical properties of weld metal.](image)
were developed by the subsequent passes. Again it can be confirmed that the cracks developed in transverse direction are all within the region of columnar grains between temper band and fusion boundary. Some of the long cracks stop their propagation at the fusion boundary and also were arrested near tempered band in the other side of crack tip. No cracks were present in the regions of recrystallized HAZ-WM. In order to find out any effect of microhardness on the formation of cold cracking, a microhardness measurement was performed in the transverse direction on this surface. In this time, finer-scale measurement was made with a spacing of 0.5 mm to determine microhardness trends across each region more precisely. After measurement, its result was overlapped on the macrograph as shown in Fig. 10. This figure shows the hardness variation traversing from the fusion zone of final bead (11th bead), across its recrystallized HAZ-WM and tempered band formed in 10th bead and through 9th bead, to the other side of base metal.

Universally, a level of high hardness near or over 340 Hv is recorded in the regions of recrystallized HAZ-WM. Such a high hardness drops dramatically near the region of tempered band resulting in the minimum hardness in the tempered band. As it approaches to the next fusion boundary the hardness increases gradually to as-deposited hardness. Thus, the one end of the crack is located in the region of minimum hardness (tempered band) and the other end is in the maximum hardness (fusion boundary). As the cracks are laid crossing such regions of gradual increase in hardness it can be said that they have little dependence on the microhardness. If the microhardness were a major factor for cold cracking the region of recrystallized HAZ-WM would be the most susceptible site to cold cracking. However none of the cracks were found in those regions of high hardness. As the hardness does not have a strong relation with the preferential site for cold cracking the only one left to be considered to be important is a microstructural aspect. One of the clear differences in microstructure is that the cracked region has a structure composed with columnar grains while the non-cracked region has a structure of recrystal- lized grains. It certainly demonstrates that, in the present welds, the local variation of microstructure plays more important role in developing cold cracks than that of microhardness.

One of the transverse cracks, marked as ‘B’ in Fig. 9, was opened in the liquid nitrogen and the crack surface was examined under SEM. Figure 11(a) shows the whole surface of the crack in a low magnification. It is easily understood that this crack is exactly the same in shape as the region indicated with dotted line in Fig. 9(b). In a high magnification fractograph taken from the center of cold crack (location ‘C’ in Fig. 11(a)), it is clear that these cracks were developed mainly in intergranular fracture mode along the columnar grain boundaries. As the crack stops its propagation at the fusion boundaries the outer boundary of crack matches well with the fusion boundary. However, in region ‘D’ marked in Fig. 11(a), crack extends its propagation slightly into the recrystallized region of HAZ-WM and shows intergranular fracture along the boundaries of equiaxed grains (Fig. 11(c)). This fact demonstrates the columnar grains are far less resistant to cold cracking than the recrystallized grains even though they are lower in hardness. At this moment, the reason for this is not clearly understood but one may say that recrystallized grains have clearer boundaries than the as-deposited columnar grains as the former are formed newly from the latter ones. In addi-
tion, as shown in Fig. 9(a), shorter cracks tend to be located near temper band, *i.e.* in the region of columnar grains with lower hardness, but not in the region near fusion boundaries. This fact indicates that the columnar grain structure becomes more susceptible to cold cracking as it is tempered to some extent. In this study, no evidence for supporting the formation of grain boundary carbide was made but the tendency of crack initiation from near tempered bands implies the extent of tempering also plays some role in crack formation.

Based on the above discussions, it was suggested that either the decrease of carbon content or the grain boundary decoration with grain boundary ferrite would be beneficial in improving the cold crack resistance of high strength weld metal. Low carbon content will suppress the unnecessarily high hardness in the upper layers as well as the possible formation of carbides along the columnar grain boundaries. The grain boundary decoration to some extent may help the weld metal increase the cold crack resistance by preventing the intergranular fracture. A further understanding of the microstructural constituents formed in high strength weld metal and their influence on cold crack susceptibility will lead to the development of preheat-free welding consumables for HSLA steels.

4. Conclusion

Using the submerged-arc multipass weld metal cracking tests with commercial welding consumables, the influence of local variation of microhardness and microstructure on weld metal cold cracking was examined. As the hydrogen also played an important role in weld metal cracking, the cracking tendency decreased with the increase of preheating temperature. Regardless of the preheating temperature, the cold cracks were found to be developed mostly in the top layer, and more importantly they appeared to be initiated from the areas having columnar grain structure that have been tempered to some extent. In contrast the heat-affected and recrystallized structure having equiaxed grains provided weld metal with better resistance to cold cracking even though they were substantially high in hardness. As a result, it can be concluded that the location of cold cracks formed in the present high strength weld metal was mainly controlled by the local variation of microstructure but little affected by the microhardness variation. Therefore the susceptibility of multipass weld metal should be different from location to location due to the microstructural variation developed during multipass welding procedure, and the most susceptible microstructure was found to be the columnar grain structure tempered to some extent. It has been further demonstrated that a localized increase in microhardness was not necessarily related to an increase in the susceptibility to cold cracking.

REFERENCES